

On secondary hardening of structural steels.

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was measured before and after tempering on the line (200); the absolute error amounted to 0.0026 kX. The experimental results are entered in tables and graphs. It was found that secondary hardening is linked with a reduction of the carbon content and possibly also of alloying elements in the residual austenite during tempering at 500 to 550°C. A qualitative correspondence was observed between the degree of drop of the carbon concentration in the residual austenite and the intensity of secondary hardening. In structural steels a preliminary condition of secondary hardening during tempering is the occurrence of partial transformation of austenite in the intermediate range during hardening. Such hardening leads to a considerable increase in the carbon concentration of the residual austenite and, therefore, during subsequent tempering at 500 to 550°C a carbide phase may separate out from the residual austenite which will result in a reduction of the lattice period of the residual austenite. In the case of steels with a higher carbon content intermediate transformation during hardening is not a necessary condition for the separation of a carbide phase during tempering at 500 to 550°C. It is possible that

Card 3/4 relaxation processes taking place during tempering play

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an important role in secondary hardening.
There are 5 figures, 3 tables and 2 Slavic references.

SUBMITTED: July 24, 1956.

ASSOCIATION: Institute of Metal Technology and Metal Physics
TsNIICHM.
(Institut Metallovedeniya i Fiziki Metallov TsNIICHM).

AVAILABLE: Library of Congress.

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ENTIN, R. I.

129-58-5-16/17

Scientific-Technical Conference on Hardening in Hot Media and Intermediate Transformation of Austenite (Yaroslavl') - 1958

by high temperature tempering;

13) A full and even a partial decomposition of the austenite in the upper region of the intermediate range causes appearance of a particular variant of irreversible temper brittleness which is characterised by a trans-crystalline fracture.

Doctor of Technical Sciences R. I. Entin and L. I. Kogan in their paper "On the Theory of Intermediate Transformation of Austenite" communicated experimental data on the elementary reactions, structure and composition of transformation products of austenite in the medium range. They pointed out that transformation in this range is not due to redistribution of the alloying elements in the austenite but to diffusional redistribution of carbon in the austenite. Depending on the composition of the steel and the transformation temperature, an increase or a decrease of the carbon concentration in the residual austenite may take place, which is due to separating out of carbides. In some cases (for instance in nickel steels) the process of carbon enrichment of the residual austenite at a later stage of the transformation is followed by a

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ENTIN, R. I., Dr. Tech. Sci.,

"Transformation of Austenite in the Medium Temperature Range," with Kogan, L. I.,
page 161.

In book Problems of Physical Metallurgy, Moscow, Metallurgizdat, 1958, 603p.
(Its: Sbornik trudov, v. 5)

The articles in the book present results of investigations conducted by the
issuing body, Inst. of Physical Metallurgy, a part of the Cent. Sci. Res. Inst. of
Ferrous Metallurgy, located in Dnepropetrovsk. The investigations were concerned
with phase transformations in alloys, strengthening and softening processes,
diffusion processes (studied with the aid of radioactive isotopes), and certain
other questions.

SOV/137-58-8-17683

Translation from: Referativnyy zhurnal, Metallurgiya, 1958, Nr 8, p 212 (USSR)

AUTHORS: Kogan, L. I., Entin, R. I.

TITLE: Transformation of Austenite in the Central Region (Preobrazheniye austenita v sredney oblasti)

PERIODICAL: Sb. tr. In-t metallov, i fiz. metallov Tsent. n. i. in ta chernoy metallurgii, 1958, Vol 5, pp 161-209

ABSTRACT: A survey of works dealing with the process of austenite (A) transformation in the intermediate region. The authors present the results of experimental investigations of processes of redistribution of C, the effect of partial intermediate transformation of A (ITA) on the kinetics of transformation during subsequent cooling, as well as the results of investigations dealing with changes in C concentration in the α phase during ITA. The investigations were performed on experimental smeltings of Mn, Cr, Si, Ni, V, Cr-Ni, Cr-Si-Mn, and carbon steels, as well as on steels alloyed with Al. It is shown that ITA is connected with a redistribution of C and with the martensite $\gamma \rightarrow \alpha$ transformation. By means of direct measurements of the period of the crystal lattice of the austenite A

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during the process of transformation, it was established that the redistribution of C in A precedes the process of the $\gamma \rightarrow \alpha$ transformation. The appearance of A zones with decreased C concentration makes it possible for martensite transformation process to occur at temperatures $>M_n$; A zones with increased C concentrations may precipitate cementite. Alloying of steel with Si significantly increases the C content in residual A. The formation of a relief in the course of ITA is an indication of martensitic character of the $\gamma \rightarrow \alpha$ transformation. The fact that certain steels exhibit a protracted incubation period and the possibility of suppressing the process of ITA during rapid cooling indicate that both these factors are connected with processes of diffusional preparation and redistribution of C in A. The effect of alloying elements on the kinetics of the ITA is determined by the manner in which they affect the diffusion rate and the diffusion length of C. The isolation of the intermediate region on the diagram of A transformation is determined by a delay in the processes of formation of carbides and polymorphous $\gamma \rightarrow \alpha$ transformation in the pearlite region, whenever steel is alloyed with such elements as Cr, Cr and Ni, Mo, W, etc. As the temperature is reduced, the rates of diffusion of the alloying elements and of selfdiffusion of Fe in the A become very small and, under these conditions, the process of transformation consists in diffusional

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redistribution of C only in conjunction with martensite $\gamma \rightarrow \alpha$ transformation in regions of A with reduced C concentration. The formation of relief and a change in the period of the lattice of the retained A. at the temperatures of upper and lower regions of the ITA, justify the assumption that the mechanism of transformation in the various regions is, basically, identical.

M. Sh.

1. Austenite—Transformations 2. Carbon—
Diffusion 3. Austenite—Structural analysis

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AUTHORS: L.I. Kogan and R.I. Entin

SOV/24-58-6-3/35

TITLE: On the Mechanism of the Isothermal Transformation of Austenite in the Intermediate Temperature Range (O mekhanizme promezhutochnogo prevrashcheniya austenita)

PERIODICAL: Izvestiya akademii nauk SSSR, Otdeleniye tekhnicheskikh nauk, 1958, Nr 6, pp 12-18 (USSR)

ABSTRACT: The authors review the results of recent investigations into those transformations of austenite which result in the formation of bainite. They reject Cottrel's findings (Ref 6) that martensite is not formed during these reactions, and conclude that the bainite transformation proceeds in two stages; diffusion and redistribution of the carbon content occur during the first stage, followed by a martensitic $\gamma - \alpha$ transformation. This hypothesis had been advanced by the authors in an earlier paper (Ref 3) describing the results of an investigation of the isothermal transformation in steels whose composition and code numbers are given in Table 1. Some results of that research are reproduced in Figs 1 to 4, in the form of graphs showing the kinetics of the transformation (the time

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dependence of the proportion of decomposed austenite) and the corresponding variation of the lattice parameter. However, no general laws could be formulated on the basis of the previously obtained data, and the present work was intended to verify the hypothesis that carbon redistribution did in fact precede martensitic transformation. This confirmation was to be achieved by measurement of the carbon content of the resultant α or ferrite phase, which should, in accordance with the theory, have a carbon content lower than the average of the investigated steel. In order to avoid confusion due to the possible decomposition of the α phase during the isothermal transformation, it was necessary to select steels in which complete transformation occurred in the intermediate temperature range, and in which martensite did not decompose below 400-450°C. Consequently, steels 24S2G3N2F and 23S2G3Kh2N2F composition which were thought to possess these characteristics were used in the present investigation. The compositions of these steels were respectively:

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0.24% C, 2.10% Si, 2.94% Mn, 1.85% Ni, 1.64% V;
 0.23% C, 2.07% Si, 2.96% Mn, 1.83% Cr, 1.85% Ni, 1.54% V.
 The experimental specimens were investigated by hardness measurements, and by magnetometric and X-ray diffraction techniques. Samples of forged material were chromium plated after vacuum homogenisation. They were then heated in argon to 1150°C, held at this temperature for 4 to 6 minutes and then transferred to the isothermal transformation bath which was maintained at temperatures ranging from 150 to 350 C for a period long enough to ensure the maximum possible transformation at the given temperature. The specimens were then quenched to room temperature in a 10% solution of caustic soda. The magnetometric measurements showed that the austenite did not transform isothermally at temperatures higher than 400°C. The T.T.T. curves for the two investigated steels are shown on Figs 5a and 6a. As can be seen from the graphs reproduced on Figs 5b and 6b, the proportion of the decomposed austenite at 350, 300 and 250°C was 5, 70 and 80%

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respectively. The carbon content of the α phase was determined by an X-ray method developed by G.V. Kurdymov et alii, E.Z. Kaminskiy and T.I. Stelletskaya (Refs 8, 12,13), which involved measurement of the width of the line diffracted from the 211 planes. The variation of the width ω (in 10^{-3} radians) and of Rockwell hardness number (R_C) of the studied steels, quenched and tempered for 1 hr at 200 to 650°C, is entered in Table 2. The proportion of the transformed austenite (%), hardness (R_C), the width of the (211) lines, and the estimated carbon content of the α phase of steel specimens tempered for 30 minutes at various temperatures are given in Table 3. The data reproduced in Table 2 indicated that decomposition of the martensite did not occur when the quenched specimens were tempered for one hour at temperatures up to 500°C. Above this temperature a decrease of hardness accompanied a decrease in width of the (211) diffraction lines of the α phase. These findings confirmed the suitability of the selected steels for the purposes of the present investigation, because they showed that no decomposition of the

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α phase would occur during isothermal transformation experiments all of which were carried at temperatures below 350°C (Table 3). A direct determination of the carbon content of the α phase in the course of the transformation thus became possible. The mean width of the (211) diffraction line, determined planimetrically, was found to decrease from $35 \cdot 10^{-3}$ radians in the quenched specimens to $27 \cdot 10^{-3}$ radians after isothermal transformation. These widths were related to the carbon content by a series of corresponding line-width determinations made on the martensite of steels containing the same proportions of the alloying elements as the investigated steels, and having carbon contents ranging from 0.05 to 0.20% (Fig 7). It was found that the carbon content of the α phase formed during isothermal transformation varied between 0.15 and 0.16%. The average carbon content of the studied steels was 0.23 to 0.24%. This decrease of 30% in the carbon content was considered to be highly significant in view of a maximum possible experimental error of only 4%. These findings were accepted as confirmation of

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the hypothesis that at the instant of its formation during the intermediate temperature transformation the carbon content of the α phase was already lower than the average carbon content of the steel. Consequently, isothermal transformation of austenite must be preceded by carbon redistribution, which leads to the formation of domains of high and low carbon concentration; the domains of lowest carbon content are least stable, and transform into martensite. The stability of the other domains will vary with the carbon content, and any tendency to reject cementite at various rates will depend upon the temperature and the content of the alloying elements.

Card 6/7 There are 6 graphs, 3 tables, and 13 references of which 11 are Soviet, 1 English and 1 German.

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On the Mechanism of the Isothermal Transformation of Austenite in
the Intermediate Temperature Range

ASSOCIATION: Institut Metallovedeniya i fiziki metallov
(Institute of Metals Technology and Metal Physics)

SUBMITTED: September 10, 1957

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80V/122-58-12-20/32

AUTHORS: Shepelyakovskiy K.Z., ~~Candidate of Technical Sciences, Entin, R.I.,~~
~~Doctor of Technical Sciences~~

TITLE: A New Method of Surface Contour Hardening Medium-Module
Gears (Novyy metod poverkhnostnoy konturnoy zakalki
shesteren srednego modulya)

PERIODICAL: Vestnik Mashinostroyeniya, 1958³⁴, Nr 12, pp 53-58 (USSR)

ABSTRACT: Surface hardening of medium module gears made from medium carbon steel carried out by induction heating usually leads to the comparatively small teeth being hardened right through, unless induction heating is performed by the "two-frequency" method. With module 4 to 6 mm, the tooth chord is 6 to 10 mm and normal induction heating hardens the teeth right through - this method is satisfactory for gears which work without shock. For auto or tractor gears, through hardening is not satisfactory, and the hardness of the core of the teeth should not exceed 46-48 Rockwell C. With "two-frequency" heating, preliminary heating to 680° C is carried out at 1000 to 10000 c.p.s. and this is followed by final heating to hardening temperature at 150,000 to 400,000 c.p.s. for 0.6 to 1 second. This method gives an evenly distributed

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hardened zone around the periphery of the teeth, and a relatively soft but tough core, the hardness of which is controlled by preliminary heat treatment. "Two-frequency" heating demands high power and is expensive (800-1100 Kw).. The authors have worked out a method of through heating gears to the hardening temperature and then quenching the surface at or about the critical rate. Since gear teeth require surface hardness of 56-62 Rockwell C, medium carbon steels could not be used if the core hardness was kept down to 45 Rc by a slow quench. Special steels were developed with low hardenability which could be quenched at a high rate with water and gave limited depth of hardness. These steels require fine Austenitic grain, limitation of the elements which encourage grain growth i.e. Mn, Cr, and Ni, and preferably addition of elements such as Ti or V which form barely soluble carbides and encourage the growth of Pearlite. The composition of three alloys with low hardenability is shown in Table 2. Figs 2,3 and 4 show hardness distribution through 12.5 mm diameter specimens for these steels, respectively, hardened

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after heating to 850°C, curves a) and to 900°C, curves b). Full lines are for samples without addition of Ti, and dotted lines with Ti added in the proportions given in Table 2. Comparison of these curves shows that reduction of Mn from 0.45 to 0.13% sharply reduces depth of hardening. Addition of Ti makes up for the effects which low Mn content would normally produce on the foregoing capabilities of the material. Table 3 gives further data on hardness at tooth surface and at its core, also the depths hardened above 55 Rc and above 45 Rc, for the three alloys given above. Table 1 shows similar measurements on specimens from normal medium carbon steel hardened under the same conditions. Gears with 4.23 module (tooth chord 6.5 mm) induction heated to 850°-900°C from special steel with 0.5 - 0.6 C, 0.1% - 0.3 Ti, and less than 0.2 Mn, less than .15 Cr, and less than 0.25 Ni, gave surface hardness 60 - 64 Rc, with depth below surface above 55 Rc from 1 to 2 mm and with core hardness 33 - 38 Rc. With normal 0.45 medium carbon steel, the surface hardness following similar treatment was 60 - 62 Rc, and the core

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hardness 55 - 58 Rc. Loads (kg) to break these teeth are given in Table 4; here the top group of figures is for normal carbon steel and the lower group for experimental steel. To enable the quenching rate to be controlled water must be sprayed at a set pressure. The induction heating and quenching rig shown in Fig 5 was constructed. The quench is controlled by relay operated magnetic valves at fixed pressure and temperature. Normal heating time to 850°C is 20 to 30 seconds. The hardened gears are tempered at 150°C for one and a half hours. Fig 7 shows a plot of the hardness over the area of a 4.23 module tooth so hardened. Fig 8 depicts the microstructure of the hardened layer at the surface, of the core, and of the material before hardening. There are 8 figures, 4 tables and 10 references (7 Soviet,

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SOV/129-59-6-2/15

AUTHORS: Kogan, L.I., (Cand.Tech.Sci.), and
Entin, R.I. (Dr. Tech. Sci., Professor)

TITLE: Certain Laws Governing the Transformation of
Residual Austenite (Nekotoryye zakonomernosti
prevrashcheniy ostatochnogo austenita)

PERIODICAL: Metallovedeniye i termicheskaya obrabotka metallov,
1959, Nr 6, pp 7-13 (USSR)

ABSTRACT: The authors investigated in detail the influence of partial intermediate transformation of austenite on the subsequent transformation at lower temperatures for the steels 53KhN3 (0.53% C; 1.1% Cr and 3.4% Ni), 60S2 (0.6% C and 2.3% Si), and 129G2 (1.29% C and 2.7% Mn). For each of these, two temperatures (T_1) of preliminary transformation were selected and two temperatures of subsequent transformation (T_2) in the minimum range. These temperatures were as follows: $T_1 = 405$ and 365 °C, $T_2 = 300$ and 260 °C for the steel 53KhN3; $T_1 = 400$ and 350 °C, $T_2 = 350$ and 300 °C for the steel 60S2. After preliminary transformation at the temperature T_1 the subsequent transformation at T_2 began after a certain incubation period. In Figs 1-3 the influence is

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graphed of the partial transformation of the austenite on the kinetics of subsequent transformation for the three tested steels. A drop in the degree of transformation at the temperature T_2 after partial transformation at the temperature T_1 was observed for all the investigated steels, irrespective of the sense of the concentration changes occurring in the austenite. This drop is particularly large if the transformation at the temperature T_1 proceeds until damping occurs. It can be assumed that this effect, as well as self-braking of the intermediate transformation at a constant temperature is related to the martensite mechanism of gamma-alpha transformation in the intermediate range. The kinetics of transformation of the residual austenite during tempering of hardened steel comply basically with the same relations which were established for isothermal austenite transformation by Cohen (Ref 11) and by the authors of this paper (Ref 12). This assumption is confirmed by the transformation diagrams of the primary and residual austenite of the steel 73KhN3 containing 0.73% C, 0.8% Cr, 3.5% Ni (Fig 4). The influence of

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partial intermediate transformation of austenite on subsequent transformation at lower temperatures is one of the basic causes of the difference in the kinetics of transformation of austenite under isothermal conditions, which also applies to continuous cooling. Behaviour of the residual austenite during tempering depends to a considerable extent on the conditions of hardening of the steel and on the chemical composition. For evolving rational heat treatment regimes it is also necessary to take into consideration the features of the kinetics of transformation of residual austenite during tempering. Preliminary experiments have shown that if steel which contains a considerable quantity of residual austenite is subjected to heat treatment, the impact strength can be considerably improved by double tempering. For this purpose it is advisable to first temper the steel at temperatures corresponding to the lower part of the intermediate range and then to increase the tempering temperature to 600 - 650 °C. In the case of direct

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tempering at 600 - 650 °C, the impact strength will be
lower. I.V. Geyzer and L.I. Volosevich participated in
the experiments.
There are 8 figures, 1 table and 12 references, 7 of
which are Soviet, 4 English and 1 German.

ASSOCIATION: TsNIChM

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AUTHORS: Sarrak, V.I., and Entin, R.I. (Moscow)

TITLE: Irreversible Temper Brittleness^{ab} in Constructional Steels

PERIODICAL: Izvestiya Akademii nauk SSSR, Otdeleniye tekhnicheskikh nauk, Metallurgiya i toplivo, 1959, Nr 6, pp 73-82 (USSR)

ABSTRACT: The object of the present investigation was to study the connection between the temper brittleness in certain steels and the process of decomposition of martensite. The experimental materials comprised several constructional, alloy steels (20KhG, 20S2G3, 20KhN4, 07KhN5M, 17KhN5M, 35S2G3, 55S2G3, 50KhG, 37KhN4SV) and two alloy irons (02KhN4 and 02Kh7). An induction furnace was used for melting, and the ingots (25 kg) were forged at 1150 °C into 14 x 14 mm rods, from which standard, notched-bar test pieces for impact tests were prepared, the notches being cut after the heat treatment. All specimens were oil-quenched from 1100 °C, and then cooled in liquid nitrogen in order to reduce to minimum the proportion of the retained austenite. After tempering at various temperatures in the 200-650 °C temperature range, hardness, impact strength, coercivity, and the width of the X-ray diffraction lines

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(lines (211) of the α -phase) which indicated the proportion of decomposed martensite, were measured. The impact tests were carried out at temperatures determined for each alloy by a series of preliminary tests conducted at various temperatures on specimens tempered at 200 °C; the temperature at which lowest impact strength was obtained was chosen as the test temperature for that particular alloy. Electron microscopy was also used for studying the structure of the experimental alloys while the character of fracture was studied with the aid of an optical microscope, after chromium-plating the fractured surfaces in order to facilitate their examination. Some of the obtained results are reproduced in Fig 1, where the impact strength (α_k , kg/cm²), width of the X-ray diffraction lines ($B \times 10^{-3}$, rad), coercivity (H_C , Oe) and Rockwell hardness (R_C) are plotted against the tempering temperature (°C) for steels 20KhG (Fig 1a) and 2082G3 (Fig 1b). All the investigated materials were found subject to temper brittleness, the critical tempering temperature range varying with the nature of the alloying additions. Thus, the critical tempering

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temperature for steels 20S2G3, 35S2G3, 55S2G3, 37KhN4SV, was higher than that for steel 20KhG; this is in agreement with the published data (Ref 5) according to which additions of silicon and particularly silicon plus tungsten, raise the critical tempering temperature. It was found that the onset of temper brittleness does not coincide with the beginning of the process of decomposition of martensite; thus, for instance, decomposition of martensite in steels 37KhN4SV, 35S2G3, and 53S2G3 begins at 200 °C, while temper brittleness does not develop below 300 °C. Generally speaking, the decrease in the width of the X-ray diffraction lines is accompanied by a decrease in the coercivity of the material; this is particularly noticeable in steel 20S2G3 (Fig 16) in which there is a simultaneous, sharp decrease in the width of the X-ray diffraction lines, coercivity, hardness, and impact strength. Owing to the presence of silicon and the low carbon content, decomposition of martensite in this steel begins at a temperature above 350 °C. In the next series of experiments, the kinetics of the onset of temper brittleness was studied. ✓

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The results for steel 20S2G3 are reproduced in Fig 2, where R_C , B , and α_k are plotted against time (h) at the tempering temperature, the experimental points marked by circles, triangles and squares corresponding to tempering temperatures of 300, 350 and 400 °C, respectively. It will be seen that in the case of this steel, decreasing the rate of decomposition of martensite brought about a deceleration of the onset of temper brittleness; thus, after 20 hours at 300 °C, the width of the X-ray diffraction lines remained practically unchanged, and so did the impact strength. When the duration of tempering was increased, both the width of the X-ray diffraction lines and impact strength gradually decreased. The significant fact is that while steel 20KhN4 is subject to temper brittleness, this weakness does not develop in iron, containing the same alloying additions and tempered under the same conditions. It is true that the impact strength of the 02Kh7 iron decreased after tempering at 400 °C, this effect being accompanied by a slight increase in hardness; however, the decrease in the impact strength in this case is a result of

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precipitation hardening (secondary hardness) due to decomposition of the α -phase and subsequent formation of finely dispersed carbides whose coalescence is inhibited, owing to the high chromium content. In the case of the 02KhN¹⁴ iron, in which (owing to its lower chromium content) no secondary hardening takes place, no decrease in the impact strength after tempering was observed. Examination of the fracture of specimens of the 20KhG steel, embrittled by tempering at 300 or 350°C and tested for impact strength at room temperature, showed that in this case brittle fracture takes place predominantly along the grain boundaries, this applying particularly to specimens tempered at 350 °C (Fig 3). Specimens of the same steel, tempered at 250 °C (i.e. below the critical tempering temperature) and tested at -90 °C, failed; in a brittle manner, across the grains (Fig 4). Electron microscope examination of the structure of steel 20KhG, tempered at 300 °C, revealed that the carbide phase is precipitated mainly at the grain boundaries; this effect is particularly noticeable in specimens tempered at 350 °C, in which carbides,

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precipitated at the boundaries of the martensite crystallites as well as at the boundaries of the original austenitic grains, formed in some cases continuous intergranular layers (Fig 5). (A similar effect was observed in specimens tempered at 400 °C in which, in addition to carbide particles precipitated in the interior of the grain, there were large agglomerates of carbides at the boundaries of the austenitic grains and martensite crystallites). Since proneness to temper brittleness can be revealed also by tensile tests, conducted under conditions inhibiting the plastic deformation, a series of room temperature tensile tests was carried out on grooved specimens of steels 20KhG and 20S2G3 (diameter - 10 mm, depth of the groove - 2.25 mm, root radius - 0.34 mm) that had been preliminarily tempered at various temperatures. It was found that the decrease in the impact strength due to brittleness is accompanied by a decrease in U.T.S. of the grooved specimens; thus, U.T.S. of steel 20KhG tempered at 350 °C (i.e. within the critical tempering temperature range) was 170 kg/mm² against 230 kg/mm² for specimens

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tempered at 200 °C (i.e. below the critical temperature, and 220 kg/mm² for material tempered at 400 °C (i.e. above the critical temperature); similar results were obtained for steel 2082G3. In the next series of experiments, the effect of small additions of aluminium and boron was investigated. The experimental materials were prepared by melting steel 20KhG, pouring part of the melt, and then adding to the remaining portion of the melt 0.15-0.30% Al or 0.002-0.003% B; thus, it was ensured that the materials prepared in this manner were similar in all respects except for the presence or absence of aluminium or boron. The results of comparative tests are reproduced in Fig 6, where the impact strength of the 20KhG steel, Al-free (curves 2 and 4) and containing 0.15% Al (curves 1 and 3), tempered at 250 °C (curves 1 and 2) or 350 °C (curves 3 and 4) is plotted against the test temperature. Fig 7 shows similar curves for steel 20KhG, containing 0.0018% B and tempered at 200, 300, 350, 425 and 475 °C (curves 1 to 5, respectively). It will be seen that addition of aluminium or boron lowers the temperature

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of cold-shortness (the transition temperature) of steel 20KhG, tempered at low temperatures. However, these additions do not prevent the onset of temper brittleness, which can be revealed by impact strength at sub-zero temperatures. Since it was found that the size of the austenitic grains in Al-bearing steel was smaller than that in the Al-free material, the obtained results confirmed the hypothesis (Ref 14) that the beneficial effect of aluminium addition is due to its effect on the size of the austenitic grains. This does not apply in the case of the B-bearing steels, in which the size of the austenitic grains was slightly higher than that in the B-free material, and the present authors offer the following explanation of the beneficial effect of this element. When steel 20KhG is oil-quenched, a partial decomposition of the unstable austenite takes place in the upper zone of the intermediate transformation range; in the presence of boron, which increases the stability of austenite, this partial transformation does not take place; since the products of the transformation in the upper zone of the intermediate transformation range

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become cold-short at higher temperatures than a material characterized by martensitic structure, introduction of boron should lower the upper temperature limit of the cold-shortness range. Isothermal transformation curves revealed that addition of boron increases the incubating period of the transformation in the pearlitic and intermediate ranges; the data provided by these curves were used to determine the heat treatment of specimens employed in experiments carried out to check the hypothesis outlines in the preceding paragraph. The specimens of a B-bearing steel were heated to 1100 °C and then transferred for 20 sec to a salt bath, maintained at 400 °C (whereby transformation of 5% austenite was attained), after which they were water-quenched, cooled in liquid nitrogen, and tempered at 250 °C for 1 hour. Impact tests carried out at various temperatures revealed that identical impact strength/temperature curves were obtained for the B-bearing specimens heat-treated in the manner described above and for the B-free material, oil-quenched and tempered at 250 °C. It is stated, in the conclusions, that since the temper brittleness is closely

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Irreversible Temper Brittleness in Constructional Steels

associated with the process of decomposition of martensite, it is not possible completely to prevent the development of this weakness. However, the data published in the literature (Refs 6, 14, 28) and the results of the present investigation indicate several ways of minimizing the harmful effect of tempering on the mechanical properties of steels. This can be achieved by 1) lowering the threshold temperature of cold-shortness of steels tempered at low temperatures (by grain-refining and lowering the content of solid and gaseous impurities); 2) shifting the critical tempering temperature range towards higher temperatures by means of alloying additions which inhibit the decomposition of martensite; and 3) ensuring more uniform decomposition of martensite.

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There are 7 figures and 28 references, of which 19 are Soviet and 9 English.

SUBMITTED: August 6, 1959

AUTHORS: Sarrak, V. I., Entin, R. I.

SOV/20-127-2-20/70

TITLE: On Relaxation-processes During the Recreation and the Deep Annealing of Hardened Steel

PERIODICAL: Doklady Akademii nauk SSSR, 1959, Vol 127, Nr 2, pp 306-309 (USSR)

ABSTRACT: Hardened samples of three steel alloys were annealed at different temperatures for one hour, and changes in the limit of resistance were then investigated. The results of these tests are given in table 1. The effect of annealing of hardened samples on the width of the X-ray interference bands is also investigated. Relevant data are given in table 2. From the fact that the width of interference bands decreases but slightly even when samples are subjected to higher annealing, it is concluded that the bands are not connected with the breaking down of the martensite structure. However, the decrease in width of the interference bands is explained by the existence of elastic deformations in the micro-regions (tensions of the second kind). As may be seen from the diagram (Fig 1) the resistance at 20°C reaches a limit 5 to 6 hours after hardening. The same is valid

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On Relaxation-processes During the Recreation and the Deep Annealing of Hardened Steel SOV/20-127-2-20/70

for internal friction. Further, changes in the internal friction dependent on annealing were examined by determining the free torsional vibrations. Results are listed in a diagram (Fig 2). These experiments were carried out at the Kafedra fiziki Moskovskogo instituta stali RKF-MIS (Chair of Physics in the Moscow Institute for Steel RKF-MIS). A decrease in the height of the peaks at 200°C was noted regardless of the annealing method. This observation is brought into connection with the breaking down of the martensite and with relaxation processes. Finally the dislocation-theory is discussed and it is stated that the local tensions are proportional to the accumulation of dislocations. Furthermore, studies are made of the relaxation time after deformation or after martensite transformation, and of the influence of the diffusion of atoms

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between 120 and 280°C on internal friction. There are 2 figures, 2 tables, and 12 references, 8 of which are Soviet.

ASSOCIATION: Institut metallofiziki Tsentral'nogo nauchno-issledovatel'skogo instituta chernoy metallurgii
(Institute for the Physics of Metals of the Central Scientific Research Institute for Iron-metallurgy)

PRESENTED: March 30, 1959, by G. V. Kurdyumov, Academician

SUBMITTED: March 21, 1959

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PHASE I BOOK EXPLOITATION

SOV/4165

Entin, Ruvim Iosifovich

Prevrashcheniya austenita v stal (Transformations of Austenite in Steel)
Moscow, Metallurgizdat, 1960. 252 p. 4,150 copies printed.

Ed.: Yu. A. Geller; Ed. of Publishing House: L.M. Gordon; Tech. Ed.:
A.I. Karasev.

PURPOSE: This book is intended for physical-metallurgical engineers and heat-treatment specialists of metallurgical and machine-building plants. It may also be used by students of schools for higher technical education.

COVERAGE: The book reviews basic patterns of pearlitic and intermediate range transformations of austenite, the influence of alloying elements on transformation kinetics and the structure and composition of the phases formed. Modern concepts of the mechanism of pearlitic transformation of austenite and its transformation in the intermediate region are discussed. Also considered are

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Transformations of Austenite in Steel

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principles of austenite transformation during continuous cooling and residual austenite transformation in the process of tempering. The author thanks G.V. Kurdymov, Member of the Academy of Sciences, K.P. Bunin, Yu. A. Geller and L.I. Kogan. There are 330 references (Soviet and non-Soviet).

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AUTHORS: Sarrak, V.I., Engineer and R.I. Entin, Doctor of
Technical Sciences, Professor

TITLE: Reversible Temper Brittleness

PERIODICAL: Metallovedeniye i termicheskaya obrabotka metallov,
1960, No.10, pp.14-19

TEXT: The main results of the investigation, described in the present paper, are reproduced graphically in Fig.7, where the critical temperature of cold-shortness of steel 20X17 (20KhG) is plotted against tempering temperature for specimens of steel (1) quenched in oil, (2) subjected to special heat treatment leading to precipitation of ferrite at the grain boundaries, (3) remelted in vacuum, (4) heat-treated, (5) containing an addition of 0.3% Al, and (6) containing an addition of 0.002% B. Analysis of these results, correlated with the findings of other workers, led the present authors to the following conclusions:
The main cause of reversible temper brittleness is the specific character of decomposition of martensite at the grain boundaries, where the transformation often begins earlier and proceeds at a rate faster than in the interior of the grains. Tempering at low
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and moderately high temperatures brings about (at a certain stage of the process) almost complete decomposition of martensite at the grain boundaries with the interior of the grains still constituting a super-saturated solid solution. The heterogeneous structure, formed in this way and containing domains of solid solution with different carbon contents, is the principal cause of the increased tendency of steel to fail in a brittle manner. When steel in this condition is stressed, the solid solution regions, characterized by low yield point, are subjected to a system of tri-axial stresses, in which case the ratio of the normal to tangential stress increases and the critical temperature of the cold-shortness is raised. Tempering, carried out at relatively high temperatures, does not cause cold-shortness since, under these conditions, no concentration gradient is formed in the solid solution. Another factor which promotes the development of reversible temper brittleness is the decrease in the free surface energy of the grains. The probability of intercrystalline fracture decreases with increasing magnitude of the difference $\gamma_n - \gamma_g$, where γ_n is the free surface energy of the grains and γ_g is the excess

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grain-boundary energy due to heavily distorted crystal lattice in these regions. Normally this difference is large but it decreases considerably in the presence of a decomposed solid solution formed at the grain boundaries of martensite, which leads to intercrystalline fracture as a result of the onset of reversible temper brittleness. It is possible that the beneficial effect of the introduction of small additions of aluminium or boron, thermo-mechanical treatment, and vacuum melting on the tendency of steel to develop reversible temper brittleness is associated with the effect of these factors on the surface energy of the grains in decomposed martensite. Thus, although it is not possible to prevent the onset of irreversible temper brittleness, the critical tempering temperature can be raised by addition of elements which retard the second stage of decomposition of martensite (silicon, tungsten), and the temperature at which the tempered steel exhibits its brittle characteristics can be depressed by the application of vacuum melting, thermal and mechanical treatment and /or introduction of small additions of aluminium or boron.

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There are 7 figures and 16 references: 9 Soviet, 5 English,
1 German and 1 French.

ASSOCIATION TsNIChM

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UTCHVSKIY, Lev Markovich; BERLIN, Ye.N., red.; BERLIN, Ye.N., red.izd-vs;
VAINSHTEYN, Ye.B., tekhn.red.

[Temper brittleness of steel] Otpusknaia khrupkost' stali.
Moskva, Gos.nauchno-tekhn.izd-vo lit-ry po chernoi i tsvetnoi
metallurgii, 1961. 191 p. (MIRA 14:4)
(Steel--Brittleness)

S/180/61/000/001/008/015
E073/E535

TRANSLATION

AUTHORS: Sarrak, V. I. and Entin, R. I. (Moscow)

TITLE: Brittleness of Steel in Conjunction with Secondary Hardness

PERIODICAL: Izvestiya Akademii nauk SSSR, Otdeleniye tekhnicheskikh nauk, Metallurgiya i toplivo, 1961, No.1, pp.117-118

TEXT: During tempering of steels alloyed with vanadium, molybdenum and tungsten, an increase in hardness occurs in the temperature range 500 to 600°C, which is associated with the formation of disperse carbides of these elements. The authors of this paper investigated the tendency to brittleness of such steels in conjunction with the process of decomposition of the solid solution of carbon in the α -phase. The investigations were made for the steels 10T, 1035T, 20C2F3H2X (20S2G3N2KhF), 30C2F3H2X (30S2G3N2F), 35C2F3H2X (35S2G3N2F), 40C2F3H2X (40S2G3N2F). The steel 10T was quenched from 1350°C, the remainder were quenched from 1100°C. The hardness R_C , impact strength a_k , kgm/cm², width of the X-ray interference lines B , mrad, and the coercive force H_C , Oe, were measured, Fig.1; the top graph relates to the

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steel 10T (0.1% C, 0.5% Ti) for which the impact strength was determined at +120°C, the bottom graph relates to the steel 20S2G3N2KhF (0.21% C, 2.07% Si, 3.02% Mn, 1.85% Cr, 1.83% Ni, 1.57% V). For the steels 10T and 35T the hardness decreased somewhat before reaching a maximum; for the other steels an increase in hardness was observed from 400°C onwards. The decrease in the impact strength occurred after the beginning of decomposition of the solid solution, which can be evaluated on the basis of the width of the X-ray interference lines. This is particularly pronounced for the steel 20S2G3N2KhF (Fig.1b), for which the temperature of the beginning of a drop in impact strength coincides with the temp. of decomposition of the solid solution. The character of the fracture of the steel in the brittle state, corresponding to maximum hardness, is intercrystalline. If the tempering temperature is increased to 650°C, the fracture will become transcrystalline. It is pointed out that development of the brittleness begins prior to reaching maximum hardness and is associated with processes of decomposition of the solid solution. It can be assumed that the development of brittleness is to some extent the consequence of a

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differing degree of decomposition of the solid solution along the boundaries and inside the grain. Decomposition of the solid solution at the grain boundaries usually occurs at greater speeds than inside the grain; it can be anticipated that the maximum degree of structural nonuniformity will correspond to a maximum hardness inside the grain, which is associated with rejection of disperse carbides. Particularly at this stage the process of decomposition can be completed along the grain boundaries. Thereby, inside the grain the saturated solid solution of carbon in α -iron, which is in metastable equilibrium with disperse rejections of special carbides, will still be maintained. This is evidenced by the nature of the changes in the widths of the interference lines of the α -phase at tempering temperatures of 300 to 500°C (Fig.1b). The presence in the alloy of sections of solid solutions of differing composition with differing yield points brings about brittle failure, as was shown for instance by V. A. Pavlov and M. V. Yakutovich (Ref.3) in the case of quenched steel with a ferrite network along the grain boundaries. Failure in this case will occur along the sections of the decomposed solid solution with

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a lower yield point which are distributed along the grain boundaries. A further increase of the tempering temperature leads to full decomposition of the solid solution inside the grain and elimination of the structural nonuniformities of the steel. The nature of the failure changes: fractures along the grain boundaries will cease. The impact strength of the steel can increase considerably (Fig.1a), although in some cases it will remain unchanged in spite of the considerable drop in hardness (Fig.1b). It cannot be ruled out that this phenomenon in the steel 20S2G3N2KhF is associated with the difficulty of the process of coagulation of the disperse carbides due to alloying of the steel with silicon, chromium and vanadium. Consequently, the impact strength may not increase, although the hardness of the steel will drop as a result of decomposition of the solid solution of carbon in the α -iron. Obviously, for increasing the impact strength of this steel high temperature tempering is required, which leads to coagulation of the disperse special carbides. The brittleness associated with failure along the grain boundaries is obviously a phenomenon which is associated with decomposition of solid solutions if at a certain stage the degree of

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decomposition along the grain boundaries differs from that inside the grains. Such brittleness was detected by A. Geisler and F. Keller (Ref.4) in alloys of aluminium with magnesium and silicon and by A. Guy, C. Barret and R. Mehl (Ref.5) in beryllium bronze and by A. Kramer and W. Baldwin (Ref.6) in stainless chromium-nickel austenitic steel. The irreversible temper brittleness of constructional steels is also associated with differing degrees of decomposition of the solid solution of carbon in the α -iron along the boundaries and inside the grain (Ref.7). There are 2 figures and 7 references: 3 Soviet and 4 non-Soviet.

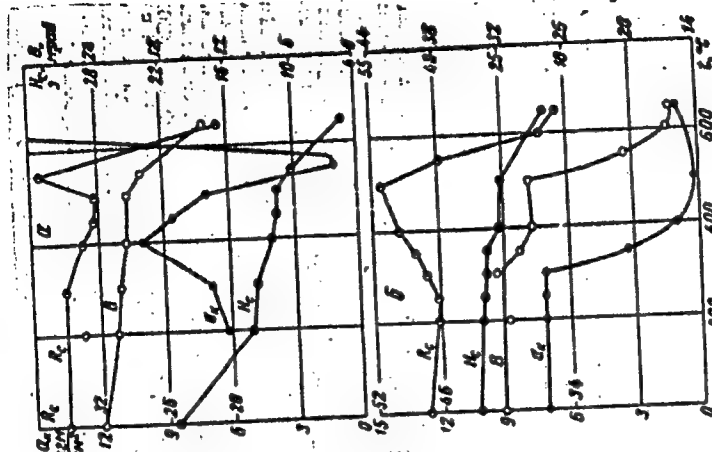
(Note: This is a complete translation)

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Fig.1



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AUTHORS: Entin, R. I., Doctor of Technical Sciences, Professor
and Kogan, L. I., Candidate of Technical Sciences

TITLE: Redistribution of Carbon During Intermediate
Austenite Transformation

PERIODICAL: Metallovedeniye i termicheskaya obrabotka metallov,
1961, No.7. pp.7-11

TEXT: According to earlier investigations of the authors,
the nature of the changes in the carbon concentration in residual
austenite depends on the chemical composition of the steel.
During transient transformation the carbon concentration in the
residual austenite becomes highly non-uniform. This follows from
metallographic investigations and from the results of measurements
of the lattice parameter of the residual austenite in the case of
step cooling to temperatures lower than room temperature.
Measured parameters of electrolytically isolated austenite also
indicate that after partial intermediate transformation the
austenite will become non-uniform. Some features of the re-
distribution of carbon in the austenite may escape notice, since

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sections of residual austenite with the lowest carbon concentration, which do not become transformed into the α -phase at the isotherm temperatures, may become transformed during the cooling process. Furthermore, the lattice parameter of the residual austenite can change, depending on the fraction of martensite in the structure. Therefore, the authors considered it advisable to determine the lattice parameters of austenite directly at the intermediate transformation temperature. In earlier work such measurements were made on steels which, after quenching, had an austenitic structure and for which the intermediate transformation occurred on heating to 300-400°C. It was found that the lattice parameter of the austenite decreases during the process of transformation. Comparison of the data of high temperature X-ray exposures with results measured after cooling the specimens to room temperature have confirmed earlier established relations and also the considerable nonuniformity of the carbon distribution in untransformed austenite. The authors used a special chamber for high temperature X-ray exposures, developed by E. Z. Kaminskiy and L. I. Kogan (one of the authors), for direct study of the austenite of various steels during intermediate transformation.

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The compositions of these steels are given herewith (in %)

Steel	C	Si	Mn	Cr	Ni	Mo
53XH30 (53KhN3S)	0.53	1.22	0.32	1.42	2.35	-
52X30.21M (52Kh3S2GM)	0.52	2.18	1.8	2.97	-	0.38
80X4 (80Kh4)	0.81	0.16	0.28	3.86	-	-

The selection of the steel compositions were based on the cooling conditions in the chamber and on the conditions of making the X-ray exposures. It was necessary to ensure a high stability of the austenite in the pearlitic range so as to prevent pearlitic transformation during cooling and also a high stability in the intermediate range so as to permit taking numerous X-ray exposures during the process of transformation. Specimens 0.8 x 10 x 100 mm were vacuum annealed for 10 hours at 1150°C and then they were etched so as to remove the decarburised layer. Following that, the specimens were heated by passage of an electric current to

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930-950°C in a helium atmosphere for a period of 5 min and then cooled to the isothermal holding temperature. The temperature was measured with an accuracy of $\pm 5^\circ\text{C}$ by means of a welded-on platinum-platinum thermocouple. The exposures were made with the K_α radiation of manganese and an exposure time of 20 min focusing onto the (311) line of the austenite. It was found that this interference line became strongly blurred after the beginning of the intermediate transformation (Table 2). Since type I distortions were virtually absent from the 0.8 mm thick specimens, which were cooled at a moderate speed, a change in the position of the centre of the line during the process of transformation must be associated with a change of the average concentration of the carbon in the austenite. In the steel 53X143 (53KhN3) the average austenite parameter increased 14-fold in the steel 52Kh3S2GM 11-fold corresponding to increases in the average carbon concentration by 0.3 and 0.25%, respectively. Another series of experiments have shown that the width of the austenite line during intermediate transformation is determined to a considerable extent by the distribution of the carbon in the

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austenite and not by type II distortions or by refining of the blocks. Investigation of the steel 80Kh4 in the medium temperature range has shown that the lattice parameter of the austenite does not change during intermediate transformation. The following conclusions were arrived at:

1. It was experimentally confirmed that a pre-requisite of $\gamma \rightarrow \alpha$ transformation in the intermediate range is the redistribution of carbon in the austenite.
2. Redistribution of carbon leads to the formation of areas with differing carbon concentrations. Areas with the lowest carbon concentration ensure the possibility of martensitic transformation at intermediate range temperatures. The peculiarities of redistribution of carbon in the austenite depend on the initial carbon content in the steel and the nature of the alloying.
3. Redistribution of carbon and presence of areas with changed lattice parameters of the austenite can be observed only if these are relatively stable and the formation of carbides is stopped. Apparently, in alloyed as well as in carbon steels containing 0.8 to 1% C, the carbides form sufficiently rapidly so that the lattice period of untransformed austenite does not change.

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There are 2 figures, 2 tables and 7 references: all Soviet.

ASSOCIATION: TsNIICHM

Table 2.

1. Frame No.
 2. Time from the beginning of the transformation, min
 3. Quantity of transformed austenite, %
 4. Width of the (311) line, mm
 5. Distance between the lines, mm
 6. Bragg angle, °
 7. Parameter at 300°C, a, kX
 8. Change in the carbon content, %
 9. Steel 53KhN3S, isothermal 300°C
 10. Steel 52Kh3S2GM, isothermal 310°C
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①	②	③	④	⑤	⑥	⑦	⑧
№ кадра	Время от начала преобразования в мин.	Количество времени, прошедшего аустенитизации в %	Ширина линии (311) в мм	Расстояние между линиями в мм	Угол Брегга θ	Параметр при 300° в к.А	Изменение содержания углерода в %

Сталь 53ХН3С, изотерма 300°. 9

1	2	3	4	5	6	7	8
0-20	0-10	1,4	26,3	74°48'	3,004	—	—
20-40	10-55	3,9	27,37	74°15'	3,313	0,20	—
40-60	55-78	4,0	27,75	74°05'	3,316	0,27	—
60-80	78-84	3,9	28,0	73°58'	3,318	0,31	—

Сталь 52Х3С2ГМ, изотерма 310°. 10

1	2	3	4	5	6	7	8
0-20	0-7	1,5	25,5	75°0'	3,598	—	—
20-40	7-24	3,5	26,0	74°55'	3,602	0,09	—
40-60	24-38	4,0	26,5	74°40'	3,606	0,19	—
60-80	38-49	3,8	26,5	74°40'	3,606	0,19	—
180-200	71-74	3,5	26,6	74°37'	3,607	0,21	—
300-320	80	3,8	26,9	74°29'	3,609	0,24	—

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AUTHORS: Kogan, L.I. and Entin, R.I.

TITLE: Intermediate Transformation of Austenite in Carbon Steel

PERIODICAL: Fizika metallov i metallovedeniye, 1961, Vol. 12, No. 2, pp. 204 - 207

TEXT: So far, no definite data are available on the upper temperature of the range of intermediate transformation in carbon steels. The causes that determine the features of intermediate transformation in carbon steels have also not been clarified. The authors investigated the formation of surface relief in carbon steel containing 0.9% C by direct microscopic investigation of a polished section of the specimen in high-temperature vacuum equipment. It was assumed that high-temperature heating and the use of thin (1 mm) specimens would make possible undercooling the austenite down to temperatures of the medium range and observation of the process of formation of the surface relief. The specimens were heated for 15 min

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at 1 150 °C and during this process the austenite grain grew to large dimensions (0.5 - 0.6 mm in diameter). The specimen was cooled to 540 °C, at which temperature it was held for 5 min (until transformation was completed); subsequent cooling to room temperature did not lead to formation of a relief. The test results indicate that this temperature is in the range of pearlitic transformation. Cooling from 1 150 to 530 °C led to the growth of only a very small number of crystals (a few crystals in each grain) during the first 3 seconds after the specimen reached this temperature. Further soaking at this temperature for 5 min and cooling to room temperature did not result in any additional relief formation. This indicates that partial intermediate transformation is followed directly by pearlitic transformation. If cooling proceeded after holding the specimen at 530 °C for 3 seconds an additional relief formed at the surface at lower temperatures of the intermediate range and below the martensitic point (230 °C). The data obtained make possible the plotting of an isothermal diagram of the

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Intermediate Transformation

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intermediate transformation of the austenite in carbon (0.9%) steel. Fig. 5 shows the diagram of the isothermal transformation ($^{\circ}\text{C}$, versus time in secs and min). The curves of the beginning and end of the austenite transformation were determined magnetometrically; the hatched (intermediate transformation) area was determined on the basis of observation of the relief formation. High-temperature metallographic investigation of the surface relief enabled establishing the range of intermediate transformation in steel containing 0.9% carbon. The upper temperature of this range is $530 - 540^{\circ}\text{C}$; in the temperature range $530 - 470^{\circ}\text{C}$ intermediate transformation develops only to a certain extent and pearlitic transformation will occur after this. The degree of intermediate transformation increases with decreasing temperature. Thus, it is shown that characteristic features of intermediate transformation exist in carbon steel, namely - self-braking at a constant temperature and temperature-dependence of the limit of transformation. The medium range of transformation could

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not be detected in numerous earlier investigations due to the high speed of the process of pearlitic transformation which follows intermediate transformation.

There are 5 figures and 10 references: 8 Soviet and 2 non-Soviet. The 2 English-language references quoted are: Ref. 6 - Ko, T. and Cottrell, S.A. - J. Iron and Steel Inst., 1952, 172, p. 3; Ref. 8 - Kazou Tsuya a Tetsutaro Mitsuttashi J. Mechanical Laboratory of Japan, 1955, Vol. 1, No. 2.

ASSOCIATION: Institut metallofiziki TsNIChM
(Institute of Physics of Metals, TsNIChM)

SUBMITTED: December 10, 1960 (initially)
January 17, 1961 (after revision)

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S/020/61/138/004/010/023
B104/B203

AUTHOR: Kogan, L. I and Entin, R. I.

TITLE: Studies of carbon concentration in the alpha phase in
intermediate transformation of austenite

PERIODICAL: Akademiya nauk SSSR. Doklady, v. 138, no. 4, 1961
826 - 827

TEXT: In the introduction, the authors discuss recent studies stating that the austenite transformation in the medium temperature range consists of a diffusion redistribution of carbon in austenite and a martensite transformation γ - α . The formation of part of the austenite with higher carbon content permits the martensite transformation γ - α at medium temperatures above the M_H point [Abstracter's note: M_H point not defined.]

It was shown that the redistribution of carbon is a necessary condition for the formation and growth of α crystals. Further, the authors had shown in a previous paper that the carbon concentration in the α -phase at the instant of transformation is lower than the mean carbon concentration in steel (Izv. AN SSSR, OTN, no. 6 (1958)). The method used for
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these studies did, however, not allow a determination of the carbon concentration in the α -phase on the transformation temperature in the medium range. For this reason, the authors made X-ray studies of the α -phase directly at the transformation temperature. They studied a low-alloy carbon steel of the composition 0.23 % C, 2.1 % Si, 3 % Mn, 1.8 % Ni, 1.8 % Cr. and 1.5 % V. The concentration of the α -phase was determined from the width of the (211) interference line. The specimens were heated in a chamber directly with electric current to 1150°C, held at this temperature for 5 min, and subsequently cooled down to the temperature of the isothermal. After the end of the austenite transformation determined by magnetometric measurements, the specimens were heated to 300°C; then, the X-ray pictures were taken. At this temperature, carbon is not released from the α -phase, and there is no new austenite transformation; this permitted a comparison of the line widths of the α -phase. For determining the carbon concentration in the α -phase from the width of the interference lines, the authors used a calibration curve plotted with the aid of studies of hardened steel with carbon contents of 0.05 - 0.25 %. Results are given in Table 1. As can be seen, the α -phase formed in an intermediate transformation contains less carbon than corresponds

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Studies of carbon concentration ...

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8/020/61/138/004/010/023
B104/B203

to the mean carbon content in steel. The carbon concentration in the α -phase is the lower, the higher the transformation temperature. There are 1 table und 5 Soviet-bloc references.

ASSOCIATION: Institut metallovedeniya i fiziki metallov Tsentral'nogo nauchno-issledovatel'skogo instituta metallurgii im. I. P. Bardina (Institute of Metallography and Physics of Metals of the Central Scientific Research Institute of Metallurgy imeni I. P. Bardin)

PRESENTED: November 30, 1960, by G. V. Kurdyumov, Academician

SUBMITTED: November 29, 1960

Card 3/4

1.1700 1454 1045

33459
S/129/62/000/001/001/011
E193/E383

AUTHORS: Kogan, L.I., Candidate of Technical Sciences and
Entin, R.I., Doctor of Technical Sciences, Professor

TITLE: The effect of deformation of supercooled austenite
on properties of hardened steels

PERIODICAL: Metallovedeniye i termicheskaya obrabotka metallov,
no. 1, 1962, 3 - 9

TEXT: A new method of improving the mechanical properties
of steel has been developed in recent years (L.V. Smirnov,
Ye.N. Sokolov and V.D. Sadovskiy - Trudy instituta fiziki
metallov, no. 18, 1956.. Ref. 3; E.M. Lips, H.V. Zuilen -
"Metall Progress", v.66, no. 2, 1954 .. Ref. 4), which consists
of plastic deformation of supercooled austenite followed by
conventional hardening and tempering treatment, and to which
the term "TMO" (abbreviation of "termomekhanicheskaya obrabotka")
has been ascribed in the Soviet Union. The main object of the
present investigation was to study the effect of this treatment
on the mechanical properties of steel 40X45C (40KhN5S) which
was chosen for this purpose because its austenite remained stable
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E193/E383

The effect of deformation

at relatively low temperatures (570 - 500 °C). The composition of this and other steels used by the present authors is given in Table 1, as follows:

X

Melt No.	Type of steel	Composition, %				
		C	Si	Mn	Cr	Ni
1 ^x	40XH5C (40KhN5S)	0.41	1.39	0.08	1.65	4.54
2	40KhN5S	0.40	1.4	0.07	1.65	4.55
3 ^{xx}	42XH5CMΦ (42KhN5SMF)	0.42	1.85	0.25	1.86	4.15
4 ^x	31XH5C (31KhN5S)	0.31	1.45	0.07	1.45	4.45
5 ^x	33XH5C (33KhN5S)	0.33	1.35	0.04	1.60	4.25

x denotes vacuum-melting

xx means that the melt contained 0.48% Mo and 0.25% V.

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S/129/62/000/001/001/011

E193/E383

The effect of deformation

In the first series of experiments the effect of both conventional treatment and TMO on vacuum-melted steel 40KhN5S (Melt 1) was studied. The conventional treatment consisted of oil-quenching the steel from 850 °C and tempering it for one hour at various temperatures. TMO was carried out in the following manner; the test piece was heated in a furnace or in a salt bath to 850 °C, after which it was transferred to a molten tin bath maintained at 525 °C. After it had cooled to 525 °C the test piece was deformed to 70% reduction with one or two strokes of a drop hammer; it was then immediately oil- or water-quenched, after which it was tempered for one hour at various temperatures between 200 and 650 °C (in some cases, rolling instead of forging was used to deform the austenite). The results of these tests (carried out on test pieces 1 mm in diameter) are reproduced in Fig. 2, where the yield strength (σ_s , kg/mm², graph a) and UTS (σ_b , kg/mm², graph b) of steel 40KhN5S are plotted against the tempering temperature (°C), the circles (1) and dots (2) relating to specimens treated by the conventional and TMO methods, respectively. It will be seen

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E193/E383

The effect of deformation

that a maximum increase in yield strength and UTS of steel subjected to TMO was attained in specimens tempered at 220 °C. Under these conditions, $\sigma_s = 260 \text{ kg/mm}^2$ and $\sigma_b = 330 \text{ kg/mm}^2$ were occasionally attained; in these cases, elongation and reduction of area were, respectively, 5 - 6 and 30 - 35%. The hardness of steel 40KhN5S, after TMO but before tempering, was HRC 59, i.e. 3 units higher than that of the same steel hardened by the conventional method. The improvement brought about by TMO, when applied to steel 40KhN5S, melted in air, was less pronounced; this is shown by data reproduced in Table 2. To check the effect of size of the test piece on the results of the process studied, tensile specimens, 3 mm in diameter, were used in the next series of experiments. The results are given in Table 3. The effects of other variants of TMO on the properties of steel 40KhN5S (melt 1) are shown in Table 4. Some significant results were obtained when TMO was applied to steel 42KhN5SMF (melt 3), in which secondary hardness is developed during tempering at 450 - 525 °C. It was found that high mechanical

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S/129/62/000/001/001/011
E193/E383

The effect of deformation

properties ($\sigma_b = 200 - 210 \text{ kg/mm}^2$, $\sigma_s = 190 \text{ kg/mm}^2$) imparted to this steel by TMO were retained after tempering at temperatures as high as 500°C . The results of the next series of experiments confirmed that TMO brought about a decrease in the size of the martensite grains. It was also established that this effect played an insignificant part in the increase in strength brought by TMO and that there was practically no difference between the block dimensions and the magnitude of stresses of the second type in specimens subjected to the conventional and TMO treatments. In the next series of experiments, carried out on steels 31KhN5S and 33KhN5S, it was established that the beneficial effect of TMO increased with increasing degree of plastic deformation and that deformation at 525°C brought about an increase in strength of the austenite. The effect of the carbon content in this steel on the effectiveness of TMO was also studied. It was found that whereas the increase in strength brought about by the application of TMO to an 0.05% C steel amounted to 10%, the corresponding figures for the 0.14 and 0.24% C steels were 19 and 21%, respectively. It was established also that with increasing

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S/129/62/000/001/001/011

E193/E383

The effect of deformation

C content of the austenite, its rate of work-hardening increased. This effect is illustrated in Fig. 4, showing stress

(σ , kg/mm²) versus strain (ϵ , %) diagrams for austenite in steels containing 0.05, 0.14 and 0.24% C (Curves 1-3, respectively) tested at 525 °C. It is stated in the concluding remarks that the relatively greater improvement in the mechanical properties of vacuum-melted steels subjected to TMO is associated with their high purity and the resultant high plasticity of both austenitic and martensitic structures in which, therefore, local stress concentrations leading to the formation of microcracks are less likely to arise. There are 4 figures, 8 tables and 8 references: 6 Soviet-bloc and 2 non-Soviet-bloc.

ASSOCIATION: TsNIChM

Card 6/11

S/020/62/146/004/009/015
B104/B102

AUTHORS: Sarrak, V. I., Entin, R. I.

TITLE: ~~Relation between~~ the temperature dependence of the yield point of iron and the interaction of dislocations with interstitial atoms

PERIODICAL: Akademiya nauk SSSR. Doklady, v. 146, no. 4, 1962, 810 - 812

TEXT: Wires 1 mm thick, 50 mm long, produced from commercial iron (0.042% C, 0.005 % N) were annealed at 1000°C, slowly cooled to 800°C and then cooled in air. A few specimens having a temperature of 730°C were quenched in water. The grain size of the specimens was approximately 0.03 mm. In two series of measurements the difference $\Delta\sigma$ between the yield point at -196°C (-20°C) and the stress at -20°C (-196°C) that corresponded to the final deformation of the specimens was determined using a method introduced by A. Cottrell and R. Stokes (Proc. Roy. Soc., A 233, 17 (1955)). Results: $\Delta\sigma$ decreases as the degree of deformation increases. If the temperature rises from -196°C to -20°C the resistance of the material to deformation changes more strongly than is the case when the temperature decreases in reverse direction. Hence consolidation of the material at
Card 1/3

S/020/62/146/004/009/015
B104/B102

Relation between the temperature ...

-196°C is unstable and is partly canceled when the temperature is increased. In specimens aged for one hour at 60°C the effect of the preceding deformation on $\Delta\sigma$ was completely eliminated, $\Delta\sigma$ becoming the same as in the annealed state. Hence the results obtained imply that the temperature dependence of the yield point is considerably affected by the interaction between dislocations and interstitial atoms. Those interstitial atoms that are not connected with dislocations hardly influence the increase in resistance to deformation when the temperature decreases. The upper yield point is caused by the stresses necessary for moving blocked dislocations, the lower yield point by those necessary for moving free dislocations. There are 2 figures.

ASSOCIATION: Institut metallovedeniya i fiziki metallov Tsentral'nogo nauchno-issledovatel'skogo instituta chernoy metallurgii im. I. P. Bardina (Institute of Metal Science of the Central Scientific Research Institute of Ferrous Metallurgy imeni I. P. Bardin)

PRESENTED: May 9, 1962, by G. V. Kurdjumov, Academician
Card 2/3

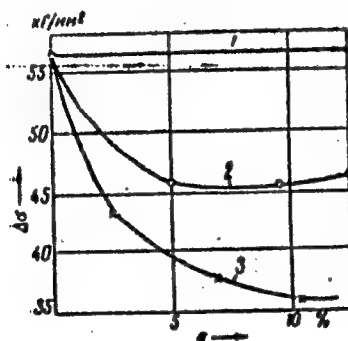
Relation between the temperature ...

S/020/62/146/004/009/015
B104/B102

SUBMITTED: May 7, 1962

Fig. 1. Effect of deformation and ageing on the increase in the resistance of iron to deformation when subjected to changes of temperature. Legend: (1) aged for 1 hr at 60°; (2) temperature change from -196°C to -20°C; (3) temperature change in reverse direction.

Fig. 1



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S/717/62/000/007/003/010
D207/D301

AUTHORS: Sarrak, V.I., and Entin, R.I., Doctor of Technical Sciences, Professor

TITLE: Brittleness of steel during tempering at low temperatures

SOURCE: Dnepropetrovsk. Institut metallovedeniya i fiziki metallov. Problemy metallovedeniya i fiziki metallov, no. 7, Moscow, 1962, 64 - 92

TEXT: The authors studied the effect of relaxation and tempering on brittleness and other mechanical properties of various steels. The steels were: 20XГ (20KhG), 20C 2Г3 (20S2G3), 20C 2Г3 H2XΦ (20S2G3N2KhF), 20C 2Г3 H2Φ (20S2G3N2F). They were quenched from 100 - 1200C in water or oil and then they were either allowed to relax at room temperature or were tempered for various periods at 100 - 65000. Brittleness, internal friction, tensile strength and hardness were measured. Immediately after quenching, the steels were brittle due to internal stresses caused by the martensitic transformation. These stresses produced elastically deformed regions where dislocations pi-
Card 1/3

S/717/62/000/007/003/010
D207/D301

Brittleness of steel during tempering ...

led up and cracks were initiated. The high internal friction immediately after quenching was due to motion of dislocations. Relaxation at room temperature and tempering at 100 or 200°C reduced brittleness and internal friction by gradual relieving of internal stresses and dispersal of dislocations. In a certain range of temperatures, lying between 200 and 500°C for different steels, tempering increased brittleness because of selective precipitation, at grain boundaries, of the solid solution of carbon in α -iron. This produced a non-uniform distribution of carbon which favored cracking along grain boundaries. Tempering at still higher temperatures produced more intense precipitation of carbon, but the distribution of the precipitate became more uniform and once again brittleness decreased. Acknowledgment is made to A.I. P'yanov, Student of the Khar'kovskiy politekhnicheskii institut im. V.I. Lenina (Khar'kov Polytechnical Institute im. V.I. Lenin) who helped in some experiments. There are 14 figures and 34 references: 17 Soviet-bloc and 17 non-Soviet-bloc. The 4 most recent references to the English-language publications read as follows: H. King and S. Glover, J. Iron and Steel Inst., 193, 123, 1959; G. Beaulien, A. Dube and G. Setendre, Trans. Met. Soc. AIME, 218, 3, 1960; R. Decker,

Card 2/3

Brittleness of steel during tempering ... 8/717/62/000/007/003/010
D207/D301

and I. Freeman, Trans. Met. Soc. AIME, 218, 277, 1960; A. Kramer, and
W. Baldwin, Trans. ASM, 50, 803, 1958.

Card 3/3

S/126/62/013/005/025/031
E111/E435

AUTHORS: Drozdov, B.Ya., Kogan, L.I., Entin, R.I.
TITLE: Influence of stress and deformation on the kinetics of
the intermediate transformation of austenite

PERIODICAL: Fizika metallov i metallovedeniye, v.13, no.5, 1962,
776-779

TEXT: Information on the effect of deformation of metastable austenite followed by quenching on the austenite transformation is incomplete. The authors have studied the kinetics of the transformation under applied-load conditions on type 40XH5C (40KhN5S) and 80X4 (80Kh4) steels. For the first, loading was carried out at 0.6 mm/min to the required stress which was then kept constant within ± 1 kg/mm². The kinetics were studied at 300 and 350°C. Acceleration occurred at all the temperatures, being especially marked at temperatures of the lower part of the intermediate region. The influence of rate of deformation was studied at 300, 400 and 525°C. This and other work shows that when conditions for thermomechanical treatment of steels are

Influence of stress ...

S/126/62/013/005/025/031
E111/E435

being worked out, the effect of stresses and plastic deformation on the austenite-transformation kinetics must be allowed for in such a way that non-martensitic transformation products should not be found. Deformation temperature should be sufficiently high to give very stable austenite, and the deformation should be as fast as possible. There are 4 figures and 2 tables.

ASSOCIATION: Institut metallovedeniya i fiziki metallov TsNIICHM
(Institute of Science of Metals and Physics of
Metals TsNIICHM)

SUBMITTED: July 25, 1961

Card 2/2

SARRAK, V.I.; ENTIN, R.I.

Studying the temperature dependence of the creep strength of iron in connection with the interaction between dislocations and interstitial atoms. Dokl. AN SSSR 146 no.4:810-812 0 '62. (MIRA 15:11)

1. Institut metallovedeniya i fiziki metallov
TSentral'nogo nauchno-issledovatel'skogo instituta
chernoy metallurgii im. I.P. Bardina. Predstavleno
akademikom G.V. Kurdyumovym.

(Creep of metals)
(Dislocations in metals)

S/133/63/000/001/011/011
A054/A126

AUTHORS: Spektor, Ya. I., Entin, R. I.

TITLE: The notch toughness and the structure of 30 XPCHA (30KhGSNA) steel after low-temperature annealing

PERIODICAL: Stal', no. 1, 1963, 87 - 88

TEXT: At the zavod "Dneprospetsstal'" ("Dneprospetsstal'" Plant) it was observed that the large sections of 30KhGSNA steel display widely differing mechanical properties and that especially the notch toughness of the steel sections varies within wide margins (2.0 - 9.0 kgm/cm²). The tests carried out (among others by magnetometry, pickling in a saturated picric acid solution at +60°C, etc.) showed the samples (hardened at 900°C, annealed at 200°C) to have a structure with strongly varying grain size. At room temperature the lower notch toughness is combined with a higher temperature limit of cold brittleness. Some samples also showed separation of ferrite at the border of austenite grains (appearing in the form of thin interlayers). (The kinetics of austenite decomposition were studied by L. I. Kogan and L. I. Yefremenko.) The presence of ferrite

Card 1/2

The notch toughness and the...

S/133/63/000/001/011/011
A054/A126

and beynite in the abnormal samples proves the insufficient stability of the overcooled austenite in the pearlite and intermediate zones. To prevent these structural deviations and to increase its notch toughness, the 30KhGSNA grade was subjected to a correcting-homogenizing annealing at 1,000°C for 15 hours. This brought back the temperature limit of cold breaking to the normal value, the stability of the overcooled austenite was increased and the grain size reduced and the notch toughness of low-temperature annealed 30KhGSNA steel improved. There are 6 figures and 1 table. ✓

ASSOCIATION: Zavod "Dneprospetsstal'" (Dneprospetsstal'" Plant) and TsNIICHM

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S/129/63/OCO/004/001/014
A004/A127

AUTHORS: Sarraf, V.I., Entin, R.I.

TITLE: Austenite hardening during thermomechanical treatment of steel

PERIODICAL: Metallovedeniye i termicheskaya obrabotka metallov, no. 4,
1963, 2 - 4

TEXT: The steel grades 45X3H8C (45Kh3N8S) - 0.45% C, 2.68% Cr, 8.35% Ni and 1.54% Si - and 40XH3M (40KhN3M) - 0.4% C, 1.76% Cr, 2.89% Ni and 0.72% Mo - were investigated to study the relation between the degree of austenite hardening and the strength of steel after thermomechanical treatment. The strength of hardened steel is determined by the level of strength of the deformed austenite and does not depend on the temperature at which the given degree of strength was obtained. If the austenite deformation temperature is reduced the coefficient of its hardening increases. The lower the deformation temperature the more considerable is the hardening of steel obtained at a lower degree of deformation. To obtain an identical hardening of steel at higher temperature, a high degree of deformation is required. It cannot be excluded that in the latter case, at the

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Austenite hardening during ...

S/129/63/000/004/001/014
A004/A127

same degree of strength after thermomechanical treatment, the steel will possess a higher ductility. There are 4 figures.

ASSOCIATION: TshIICHM

Card 2/2

SPEKTOR, Ya.I.; ENTIN, R.I.

Impact toughness and the structure of 30KhGSNA steel following low-temperature tempering. Stal' 23 no.1:87-88 Ja '63. (MIIA 16:2)

1. Dnepropetrovskiy staleplavil'nyy zavod vysokokachestvennykh i spetsial'nykh staley i Tsentral'nyy nauchno-issledovatel'skiy institut chernoy metallurgii.

(Steel alloys—Testing)(Tempering)

[illegible]

--- -- CO --- -- gas from titanium hydrogen annealing. internal diffusion,

1 21721-45

... along on dislocations. The authors data coincide with earlier investiga-
tions on the nature of strain aging in iron. Orig. art. has 4 figures.

SUB CODE: MM

NO REF SOV: 013

OTHER: 002

SUVOROVA, S.O.; SARRAK, V.I.; ENTIN, R.I.

Investigating the deformation aging of commercial-grade iron. Fiz. met.
i metalloved. 17 no.1:105-111 Ja '64. (MIRA 17:2)

1. Institut metallofiziki Tsentral'nogo nauchno-issledovatel'skogo
instituta chernoy metallurgii im. Bardina.

ACCESSION NR: AP4034537

8/0020/64/155/005/1054/1057

AUTHOR: Spektor, Ya. I.; Sarrak, V. I.; Entin, R. I.

TITLE: Explanation of the effect of nickel on cold shortness of iron

SOURCE: AN SSSR. Doklady*, v. 155, no. 5, 1964, 1054-1057

TOPIC TAGS: iron cold shortness, nickel iron alloy, plastic deformation, internal friction, ductility brittleness transition

ABSTRACT: Alloying of commercial iron and construction steels with nickel considerably lowers the temperature of transition from a ductile to a brittle state in impact tests. The authors have investigated this phenomenon on iron with 0.03% C (MZh), with and without 3.4% Ni (ON4). The heat treatment was such that the grain size was the same in both materials. Impact tests were conducted in the temperature range from -100 to -196 C. Alloying lowered the brittleness temperature by about 30 C. Since the interaction between the interstitial atoms (carbon, in this case) and dislocations which cause brittleness is reflected in the amplitude dependence of internal friction; the authors determined the latter with a torsion pendulum. The critical amplitude which was found at 20 to 200C is considerably

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ACCESSION NR: AP4034537

lower in nickel iron. This shows that the interaction of dislocations with carbon are weaker, thus lowering the fixation of dislocations and favoring plastic deformation. Orig. art. has: 4 figures.

ASSOCIATION: Institut metallovedeniya i fiziki metallov central'nogo nauchno - issledovatel'skogo instituta chernoy metallurgii im. I. P. Bardina (Institute for Metallurgy and Physics of Metals of the Institute for Ferrous Metallurgy).

SUBMITTED: 06Nov63

DATE ACQ: 13May64

ENCL: 00

SUB CODE: 111, MM

NO REF SOV: 005

OTHER: 007

Card 2/2

SUVOROVA, S.O. (Moskva); SARRAK, V.I. (Moskva); ENTIN, R.I. (Moskva)

Investigating the strain aging of iron by the internal
friction method. Izv. AN SSSR. Met. i gor. delo no.4:127-
130 J1-Ag '64. (MIRA 17:9)

KURDYUMOV, G.V., akademik; ENTIN, R.I., doktor tekhn. nauk

Some trends in the development of theoretical study of metals.
Vest. AN SSSR 34 no.10:18-32 O '64.

(MIRA 17:11)

CIA-RDP86-00513R00041212

[illegible]

CIA-RDP86-00513R000412120

"APPROVED FOR RELEASE: Thursday, July 27, 2000

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"APPROVED FOR RELEASE: Thursday, July 27, 2000

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SARRAK, V.I.; SUVOROVA, S.O.; ENTIN, R.I.

Studying the phenomenon of the deformation aging of iron. Probl. metal-
loved. 1 fiz. met. no.8;125-143 '64. (MIRA 18:7)

KOGAN, I.I.; ENTIN, R.I.

~~.....~~
Austenite transformation in the intermediate region. Probl. metalloved.
1 fiz. met. no.8;222-226 '64. (MIRA 18:7)

SARRAK, V.I. (Moskva); SUVOROVA, S.O. (Moskva); ENTIN, R.I. (Moskva)

Effect of deformation and aging on the amplitude characteristic
of internal friction in martensite. Izv. AN SSSR. Met. no.4:156.
158 JI-Ag '65. (MIRA 18:8)

ENTIN, R.I. (Moskva)

Present state of the problem of transformations in supercooled
austenite. Izv. AN SSSR. Met. no. 5:69-82 8-0 '65.

(MIRA 18:10)

BERESNEV, G.A. (Moskva); SAPRAK, V.I. (Moskva); ENTIN, R.I. (Moskva)

Effect of temperature and interstitial impurities on energy
scattering during small shifting of dislocations in iron.
Izv. AN SSSR. Met. no.6:111-119 N-D '65. (MIRA 19:1)

1. Submitted February 12, 1965.

I. 24328-66 EWT(m)/T/EWP(t) IJP(c) JD
ACC NR: AP6010425

SOURCE CODE: UR/0020/66/167/002/0322/0325⁶⁶

AUTHORS: Beresnev, G. A.; Sarrak, V. I.; Entin, R. I. ²⁵
B

ORG: Central Scientific Research Institute of Ferrous Metallurgy im.
I. P. Bardin (Tsentral'nyy nauchno-issledovatel'skiy institut
chernoy metallurgii)

TITLE: Temperature dependence of the resistance of iron to deforma-
tion and the dislocation mobility ²⁷

SOURCE: AN SSSR. Doklady, v. 167, no. 2, 1966, 322-325 ¹⁸

TOPIC TAGS: iron, temperature dependence, crystal dislocation,
crystal deformation, elastic stress, internal friction, crystal
impurity, crystal lattice

ABSTRACT: To check on the causes of the strong temperature dependence
of the elastic limit of metals with body-centered-cubic lattice, the
authors have investigated the influence exerted on this elastic limit,
taken as a function of the temperature, by the resistance of the
crystal lattice itself to the motion of dislocations. The tests were ²

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1/2

UDC: 539.377

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ACC NR: AP6010425

made on commercial iron (0.025% C, 0.005% N) containing 0.33% Ti to bind the carbon and nitrogen into carbides and nitrides of titanium. The internal friction was measured at a frequency of approximately 1 cps. The amplitude dependent internal friction was measured at strains from 2×10^{-5} to 20×10^{-5} in a longitudinal constant magnetic field (325 Oe). Increasing the interstitial impurities from 10^{-7} to $10^{-3}\%$ greatly increases the resistance to deformation. The authors describe the effects produced by interstitial impurities, by the changes in density of the free dislocations participating in the deformation, and by the motion of the free dislocation. It is concluded from the results that the temperature dependence of the elastic limit of iron is essentially the consequence of an increase in the resistance of the crystal lattice to the motion of the free dislocations with decreasing temperature. The influence of the impurities and of the density of the free dislocations comes into play to the extent that they change the dislocation velocity, and the density of the moving dislocations or the multiplication of dislocations. This report was presented by Academician G. V. Kurdyumov. Orig. art. has:

4 figures and 2 formulas.

SUB CODE: 20, // SUBM DATE: 24May65/ ORIG REF: 004/ OTH REF: 005

Card 177 2/2